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Carderock Division
Naval Surface Warfare Center
Bethesda, Md. 20084–5000

CARDEROCKDIV–SME–92/64 April 1993
Ship Materials Engineering Department
Research and Development Report

Advances in Low Carbon, High Strength Ferrous Alloys

by

E.J. Czyryca
M.G. Vassilaros

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PREFACE

The Metals and Welding Department of Carderock Division (CARDEROCKDIV), Naval Surface Warfare Center participated in the Indo–US Pacific Rim Workshop on "Advances in Low Carbon, High Strength Ferrous Alloys" held in New Delhi, India, 25 to 28 March 1992. Dr. O.P. Arora was on the Organizing Committee, Mr. E.J. Czyryca and Dr. M.G. Vassilaros gave invited presentations on the status of Navy programs on steel plate and weld metal research and development, respectively. The workshop was jointly sponsored by Office of Naval Research, Naval Research Laboratory, US Army Research Office – Far East, and the National Metallurgical Laboratory of India. The workshop included sessions on low carbon steel research given by government and industry representatives from Australia, Japan, and Korea, as well as the US and India, as follows:

* Phase Transformation & Strengthening Mechanisms;
* Structure – Property Correlation;
* Thermo-mechanical Processing & Intercritical Treatment;
* Application Areas of Low Carbon Ferrous Alloys; and
* Fluid Flow & Welding.

The objective of the symposium and workshop meetings was to organize a cooperative effort among the participants in areas of common interest for future information exchange. The workshop concluded with a discussion of on–going and planned Indo–US projects in steel research with the CARDEROCKDIV and NRL participants.

This report contains papers based on the two presentations. Paper 1 – "Advances in High Strength Steel Technology for Naval Hull Construction," by E. J. Czyryca, was presented by Mr. Czyryca. It covers an overview of US Navy steel plate research programs based on previously cleared and published technical papers. Paper 2 – "The Development of High–Strength, Cooling-Rate Insensitive Ultra-Low-Carbon Weld Metals," by M. G. Vassilaros and E. J. Czyryca, was presented by Dr. Vassilaros. It presents the status of research in high-strength steel weld metal systems based on ultra-low carbon bainitic (ULCB) metallurgy for systems in the range of 100,000 to 150,000 psi yield strength (690 to 1035 MPa). ULCB weld metals show a potential for wire electrode formulations for HSLA steel welding, where the as-deposited weld metal strength is independent of weld metal cooling rate with a good low-temperature toughness.

The Navy has maintained a high strength steel research and development program with in-house, academic, and industrial research to meet a goal of reducing shipbuilding costs. Affordability will be a primary consideration in future ship and submarine construction programs. The Navy has a vested interest in supporting research and development of high productivity, high strength steel systems and in benefiting from cooperative, international technical exchange in steel research.

ADMINISTRATIVE INFORMATION

This report presents summaries of steel plate and welding research and development programs conducted at the Center since 1980, primarily on High Strength, Low Alloy (HSLA) steels and welding of HSLA steels. The CARDEROCKDIV HSLA Steel Program has included a wide scope of RDT&E programs, ranging from 6.1 in basic metallurgy, 6.2 efforts in the alloy development of plate and welding products, to 6.3/6.4 programs to perform extensive evaluations to certify HSLA steels for ship and submarine structural applications.

The 6.1 studies were sponsored by the Office of Naval Research under Program Element 61153N managed by Dr. G. Yoder. The 6.2 investigations were conducted under both the Surface Ship and Submarine Materials (Structural) Block Programs, under Program Element 62234N, managed by Mr. I. L. Caplan, CARDEROCKDIV 0115.

The HSLA–100 programs were jointly sponsored by the SEAWOLF Acquisition Program of the Naval Sea Systems Command (NAVSEA PMS 350AT) under Program Elements 63561N and 64561N, by the Advanced Submarine Research and Development Program (NAVSEA 92R) under Program Element 63569N, and by the Aircraft Carrier Program (NAVSEA PMS 312) under Program Element 64567N.

The technical agent for most of the work described herein was Mr. C. L. Null of the Naval Sea Systems Command (SEA 05M2). The work was conducted under the supervision of Mr. G. A. Wacker, Head, Ship Materials Engineering Department (CARDEROCKDIV 28).

ACKNOWLEDGMENTS

The authors are grateful for the efforts of the many scientists, engineers, technicians, and welders of the Center's Metals and Welding Division who completed the extensive amount of planning, testing, analysis, and management of the work described herein.

ADVANCES IN HIGH STRENGTH STEEL TECHNOLOGY FOR NAVAL HULL CONSTRUCTION

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ABSTRACT

The High-Strength, Low-Alloy (HSLA) steel program was initiated by the U.S. Navy to meet objectives for the reduction of shipbuilding costs by development of more weldable steels to meet the strength and toughness requirements of HY–80 (540 MPa yield strength). HSLA–80 steel is a low carbon, copper precipitation strengthened steel based on ASTM A710 steel. HSLA–80 has been used in surface ship structural applications since 1984, after an evaluation of properties, welding, and structural performance. A substantial reduction in hull fabrication costs and higher productivity was achieved through substitution of HSLA–80 for HY–80, with the significant factor in cost savings being the reduction or elimination of preheat for welding.

Based on the HSLA–80 system, HSLA–100 steel was developed to meet or exceed the strength and toughness of HY–100 steel (690 MPa yield strength) and be weldable with reduced preheat. HSLA–100 is also a very low carbon, copper strengthened steel, based on different metallurgical principles than HSLA–80. HSLA–100 has been used since 1989 for use in surface combatant structures and ballistic protection as a replacement for HY–100 to gain significant cost reductions over a wide range of plate gages.

In addition, research programs continue to explored the properties of thermo-mechanically processed ultra-low carbon bainitic (ULCB) and accelerated cooled/direct quenched (AC/DQ) HSLA steels, with a focus on the feasibility of using these metallurgical concepts to achieve highly weldable naval shipbuilding steels with lower total alloy content and reduced dependency on heat treatment schedules. These research efforts provide an important step toward the formulation of very high strength steel systems of high fracture toughness, where properties are not restricted to limited fabrication procedures and plate gages.

KEYWORDS: metallurgy, high-strength steels, welding, mechanical properties, shipbuilding, HY–80, HY–100, HY–130, HSLA–80, HSLA–100.

INTRODUCTION

Modern warship designs in the past two decades have shown a continuing trend of increased utilization of high strength, alloy steel plate for weight reduction, increased payload, and increased mobility. Naval ship structures are subjected to a complex spectrum of dynamic loadings in service and stresses built into the hull during fabrication and fit-up. The routine dynamic loads in service include wave loadings, sea slap, slamming, vibration, thermal excursions, cargo buoyancy, aircraft/helicopter landing, and weapons reactions [1]. The ship structure must operate in both tropical and arctic seas, and, therefore must be constructed of a steel system with properties to allow reliable operation over a temperature range of about –30° to +120° F [2].

The structural integrity of the hull must be assured for continuous seakeeping in these severe environments, as well as in response to the effects of hostile weapons. The dynamic loadings, particularly in the form of shock waves, must be considered when assessing materials performance and fracture safety [3]. The fracture safety of Navy ships is addressed through the use of structural steels, and welding materials used in hull fabrication must demonstrate high fracture toughness and flaw tolerance for these extreme service conditions [4].

THE HY–SERIES NAVAL STEELS

The HY (High Yield)-series of steels have been used in U. S. naval warships since 1950. HY–80, the first in the series, was developed to achieve the optimum in high strength and fracture toughness in a welded steel system [5,6]. HY–100 and HY–130 steels were progressively developed to meet increasing strength requirements, while maintaining a high fracture toughness.

The HY–series has its origins in the quenched and tempered, Ni–Cr battleship armor steels produced by the Krupp Steel Works of Germany, circa 1894. The STS armor and pressure hull steels of the 1940's evolved from Krupp armor, and the acronym "HY–80" was applied to a special "Low Carbon STS." The HY steels are metallurgically classified as quenched and tempered martensitic steels. These steels have a martensitic microstructure resulting from the combination of alloying (Ni, Cr, Mo, and V) and the heat treatment employed to provide the optimum combination of strength and toughness. The heat treatment to develop the martensitic structure requires fast cooling from a temperature above 1400° F (austenite range); accomplished in plate, extruded, forged, or cast products by rapid water quench. The alloying elements promote martensite formation, but the element with the strongest effect in producing a martensitic structure is carbon. The as-quenched martensite has high strength and hardness, but is brittle (low toughness) and susceptible to hydrogen cracking (cold cracking). The optimization of strength and toughness is achieved by a tempering heat treatment in the range of 1000° to 1250° F.

During the welding process, the region of the joint near the fusion zone is reheated into the austenite range and cools as the welding arc continues deposition of the weld bead. If the joined sections are thick and heat is rapidly conducted away, or welding heat input is low, cooling rates in the heat-affected zone (HAZ) are high, with conditions favorable for forming untempered martensite after the weld pass, rendering it susceptible to HAZ cracking due to its high hardness. The cooling rates characteristic of arc welding plate gages over 1/2 inch are capable of producing some martensite in almost any carbon steel [7].

Cold cracking (post-weld cracking, usually below 200° F) is cracking associated with dissolved hydrogen and martensitic structure in the HAZ. Underbead cracks (parallel to the fusion line), toe cracks (propagating away from the weld toe), and delayed cracking (occurring days after welding) are different types of the same cracking mode, which originate by the same mechanism [7]:

a. Dissolved hydrogen (from shielding gas, flux, electrode coating, or surface contamination);

b. Tensile stresses (external restraint, differential expansion during welding, or transformation stress); and

c. Low ductility microstructure (untempered martensite).

One means of avoiding HAZ cracking in a welded high strength steel is a reduction of the carbon content to levels too low to produce martensite. This requires replacement of carbon strengthening with expensive alloy additions.

Preheating is an effective means of obviating both HAZ cracking and weld metal cracking. Preheating provides for the following: (1) it lowers the cooling rate to avoid martensite formation, (2) it provides time and temperature for hydrogen diffusion out of the metals, and (3) it lowers the magnitude and rate of shrinkage to reduce residual stress. The combination of a low welding heat input needed to maintain weld metal strength and the preheat needed to prevent cold cracking results in low productivity welding and high total fabrication costs for large structures.

To obtain a weldable steel meeting the strength and toughness requirements, an upper limit of 0.18 weight % was set for carbon for good weldability, and HY–80 is properly a "low carbon" alloy steel [5]. The chemical compositions of the HY–series steel plate products are given in Table 1. HY–100 is a modified version of HY–80, typically having slightly higher C, Ni, Cr, and Mo content for plate of the same gage, and tempered at a lower temperature to gain the increased strength while maintaining toughness and weldability equivalent to HY–80. The carbon content of HY–130 was limited to 0.12% maximum, compared to 0.18% maximum for HY–80, to provide adequate weldability (i.e., better HAZ cracking resistance), but provide sufficient hardenability to achieve a microstructure of predominantly tempered martensite in heavy gages. The high nickel content was needed to ensure the high toughness and survivability required in thick, welded submarine hulls at the minimum service temperature of +30° F. The chromium and molybdenum additions improved hardenability and promoted the formation of martensite in thick sections, while vanadium provided resistance to softening at the temperatures required to achieve optimum toughness. Also, low sulphur and phosphorus contents were required for improved toughness and weldability.

HY-SERIES STEEL FABRICATION

In ship hull construction, welding is the greatest cost driver and largest single shipyard labor factor. Ship structural fabrication, including materials, welding, and nondestructive evaluation (NDE), can constitute 20% of the total shipbuilding cost. The welding department in a shipyard employs twice the labor force of any other single trade.

The costs of welding include the welding consumables, the expense of preheat, electrode/flux conditioning and storage, high labor costs due to productivity limitations (heat input and welding position restrictions), necessity for repair welding, and post-weld soaking to eliminate cracking, when required. Economy demands that, in the construction of such a massive, complex structure as a ship's hull, that weldment properties be achieved "as welded."

Since the fracture toughness and transition behavior of high strength weld metals is generally inferior to that of the parent plate, overmatching weld metal yield strength is required for "protecting" the weld metal. In addition to the fracture considerations, other production welding concerns are addressed by the overmatching weld metal strength requirement. Electrode, procedure, and welder qualification weldments are fabricated under controlled conditions. These quality control items function to minimize variations among products, manufacturers, electrode lots, welders, equipment, etc. The properties of production welds may vary, depending on the particular welding conditions, and may not duplicate or even closely approach the values prescribed for test welds [8]. Also, welding in the shipyard production environment may be conducted at less than optimum conditions for achieving the properties required for the weld metal, and weld defects and discontinuities may occur, which detract from the load-carrying capacity of the weld joint. The concept of overmatching weld metal is also intended to compensate for this possible strength loss.

The welding heat input is indicative of the weld metal deposition rate for a welding process, and therefore is directly related to the cost to fabricate a structure. The limits on heat input are established by those parameters necessary to provide minimum strength and toughness, crack-free weld metal and HAZ, and operability of the welding process (the ability to make a good fusion joint). A high preheat restricts the maximum heat input in order to result in a weld metal cooling rate sufficient to achieve satisfactory weld metal mechanical properties. With a reduced heat input, more welding passes are required to fabricate a weld joint. This results in higher fabrication costs due to the labor time required to complete the structure as a result of the decrease in weld metal deposition rate. The energy costs associated with welding preheat exceeded the cost of all other energy consumed in shipyards engaged in submarine construction. Other costs and reduction in productivity associated with the mandatory use of welding preheat include: (1) capital and replacement costs for heaters, (2) labor costs for installation and removal of heaters, and (3) higher cost to remove/repair the thousands of temporary attachments employed during ship construction. Additional indirect costs associated with the preheat requirement include the disruption of other shipyard trade functions, reduced welder productivity from heat (especially in hot weather and in enclosed and/or limited-access areas), and inspection delays while the structure is allowed to cool.

A reduction in preheat temperature can increase productivity by increasing the operational envelope (range of satisfactory welding conditions) of the welding process. For example, the welding process can impart a large quantity of energy to the work piece using heat inputs up to 100 kJ/inch for HY–80 thick section applications. This energy causes the temperature of the weld joint to increase. By beginning with a lower preheat temperature, continuous welding can be maintained for a longer period of time before the maximum interpass temperature (300° F) is exceeded and the weld joint must be allowed to cool between weld passes. The maximum allowable interpass temperature of 300° F is imposed to ensure a minimum weld metal cooling rate for satisfactory mechanical properties of the weld metal. The tensile properties have been found to be particularly sensitive to the weld metal cooling rate. As the weld metal cooling rate decreases, the yield strength is found to decrease for the welding consumables employed for high yield strength steels.

HSLA–80 STEEL DEVELOPMENT AND QUALIFICATION

The HSLA steel development program identified ASTM A710 steel, a very low carbon, copper precipitation strengthened steel, as the prime candidate among a number of candidate commercial HSLA steel plate products. The evaluation program demonstrated that A710, Grade A met a minimum yield strength requirement of 80,000 psi yield strength through 3/4 inch gage, had high Charpy V-notch impact energy at low temperatures, and excellent weldability [9]. A710 steel was selected since it was in commercial production, and short-term modification could result in an easily weldable replacement for HY–80.

The A710 alloy steel was developed by the International Nickel Company to provide a field-weldable, high strength steel with good low-temperature fracture toughness for Arctic pipeline applications. Other steels considered included low-carbon, controlled–rolled and quenched and tempered HSLA pipeline steels, which typically could not meet the minimum Charpy V-notch impact toughness requirement of 35 ft-lbs at –120° F. HSLA–80 is a ferritic steel. The microstructure of the quenched and aged plate product is generally an acicular ferrite in gages less than about 1/2 inch, but polygonal ferrite in thicker plate [10,11]. Ferritic steels are widely used in civil construction because of their excellent weldability.

In 1984, HSLA–80 steel was certified for use in ship construction. Certification requires an evaluation of a structural fabrication system which demonstrates that the system will meet all aspects of structural performance equivalent to or better than the system it replaces. Material specifications and fabrication/inspection documents are based on the results of the certification program. An extensive evaluation of HSLA–80 properties, welding, and structural performance [9] demonstrated that the very low carbon, copper precipitation strengthened steel met the requirements of HY–80 steel, and was readily weldable with no preheat (32° F maximum) using the same welding consumables and processes as for HY–80 steel fabrication.

WELDABILITY OF HSLA–80 STEEL

As noted, the primary reason for preheat in the welding of the HY–series steels was to mitigate underbead cracking (hydrogen related) in the hard, martensitic heat affected zone (HAZ). The HAZ of the very low carbon, copper-strengthened HSLA–80 steel does not significantly harden, but may soften, due to the dissolution of copper and grain coarsening caused by the heat of welding [12]. HSLA steel HAZ microstructures are less sensitive to hydrogen cracking, and thus achieve excellent weldability. The very low-carbon HAZ in HSLA–80 weldments resulted in a highly weldable steel, insensitive to hydrogen cracking.

Welding process development, weldability, and shipyard producibility studies were conducted in the HSLA–80 steel certification program for the thickness range of

HSLA–80 plate from 1/4 to 3/4 inch were evaluated, and HSLA–80 steel exhibited excellent weldability for a broad range of welding conditions and processes. Overall, HSLA–80 steel was found to be superior to or at least equivalent to HY–80 steel in every aspect of ship fabrication. Welding consumables specifically for welding HSLA–80 were not developed. The welding consumables qualified for welding HY–80 steel were utilized for all welding processes. A summary of the salient features of HSLA–80 welding was as follows:

> Both hot cracking and cold cracking resistance of HSLA–80 steel, when welded with "HY–80 type" consumables, was superior to HY–80 welded under equivalent conditions.
>
> Preheat/interpass temperature and electrode handling requirements for HSLA–80 applications are less stringent than for HY–80 welding.
>
> The FCAW, SAW, SMAW, and GMAW processes may be applied satisfactorily over the whole range of welding operating conditions as long as the nominal weld metal cooling rate is 10° F/sec or higher. The requirement, the same for HY–80, assures weld metal yield strength above the minimum at high heat input levels, for thin plate, and for high preheat/interpass temperatures. The minimum preheat and interpass temperature for welding HSLA–80 was 60°F for critical applications and 32° F for general structural welding.

The relative fabricability and repairability of HSLA–80 in a shipyard production environment were established and demonstrated that significant fabrication cost reductions vice HY–80 welding could be achieved [13]. The primary factor was the cost reduction and improved productivity associated with preheat elimination. Over 20,000 plate-tons of HSLA–80 steel to MIL–S–24645 have been delivered for use in ship construction. Production has primarily been in plate gages up through 3/4 inch.

HSLA–100 STEEL DEVELOPMENT AND QUALIFICATION

Following the HSLA–80 program, an alloy development and qualification program commenced which resulted in HSLA–100 steel as a replacement for HY–100 in order to reduce fabrication costs. HSLA–100 is also a very low carbon, copper precipitation strengthened steel, based on different metallurgical principles than HSLA–80, meeting the strength and toughness of HY–100 steel, but weldable without the preheat requirements of HY–100, using the same welding consumables and processes as used in welding HY–100.

The program for the development of HSLA–100 steel consisted of three phases: (1) laboratory alloy design; (2) trial plate production; and (3) plate production for the certification program. The composition of the steel was formulated from a progressive optimization involving more than 40 laboratory-scale heats. Laboratory plates in thicknesses of 1/4, 3/4, 1 1/4, and 2 inches of HSLA–100 exceeded the minimum strength and impact toughness requirements. The results of the laboratory phase were used to develop an interim specification for HSLA–100 steel plate and as the basis for commercial production of HSLA–100 steel by steel plate mills.

The compositions of HSLA–100 and HSLA–80 steels by MIL–S–24645A, are given in Table 2, compared with the specified composition ranges for HY–80 and HY–100. The nominal compositions of HY–80, HY–100, HSLA–80, and HSLA–100 are also compared in the weldability diagram (Graville diagram), Figure 1, showing the calculated carbon equivalents for the steels [14]. HSLA–100 also has low carbon content for good weldability exhibited by steels of less than 0.10% carbon, even with significant alloying. As Table 2 shows, both HSLA–80 and HSLA–100 steels exceed the limits of "low alloy" in the HSLA acronym.

The microstructure of the quenched and aged HSLA–80 plate is generally acicular ferrite in gages less than about 1/2 inch, but polygonal ferrite in thicker plate [10,11]. Ferritic steels typically demonstrate a toughness/temperature transition curve with a steep drop in toughness over a narrow temperature range. HSLA–100 has a higher copper content than HSLA–80 for additional precipitation strengthening, and increased hardenability was achieved by increases in manganese, nickel, and molybdenum. Nickel, the greatest increase over that in HSLA–80, lowers upper shelf impact toughness, but also lowers (improves) the impact toughness transition temperature. The microstructure of HSLA–100 steel was identified by optical and scanning electron microscopy as low-carbon martensite or bainite, depending on plate gage, a significantly different metallurgy and microstructure than the ferritic HSLA–80 steel [15].

One of the benefits of HSLA–100 steel is a weldment HAZ with excellent strength and toughness. As noted for HSLA–80 steel, the HAZ of the Cu–strengthened steel does not significantly harden but may soften, due to the dissolution of copper and grain coarsening caused by the heat of welding [12]. HSLA–80 steel HAZ microstructures are less sensitive to hydrogen cracking, and thus achieve excellent weldability. The low-carbon HAZ in HSLA–100 weldments also results in a highly weldable steel, insensitive to hydrogen cracking, but not prone to soften in the HAZ, due to the increased hardenability available.

HSLA–100 STEEL PLATE PRODUCTION

An initial 150–ton commercial melt of HSLA–100 steel was produced using conventional electric furnace and ingot casting practice, conducted to achieve a very low carbon composition [16]. The minimum strength and toughness requirements of the interim specification were met in the initial production of HSLA–100 steel plate in gages from 1/4 to 2 inches. Optimum properties in HSLA–100 plate resulted from aging temperatures from 1150° to 1275° F. A double austenitization and quench was used for HSLA–100 steel plate in gages over 1 1/4 inches to refine the heavy plate grain structure for optimum toughness [15,17].

A second melt of HSLA–100 steel, again by electric furnace and ingot casting, was produced to demonstrate plate greater than 2 inches thick. The minimum strength and toughness requirements were met in plate thicknesses from 1/2 to 3 3/4 inches [16]. Subsequent production melts of HSLA–100 steel by several steel mills have produced over 5000 plate-tons of HSLA–100 steel rolled and heat treated plate to MIL–S–24645A. Melting and refining practices typically included extended decarburization to achieve a very low carbon, extra-low sulphur practice, vacuum degassing, and calcium treatment for inclusion shape control, employing argon stirring or blowing.

HSLA–100 STEEL CERTIFICATION PROGRAM

Upon receipt of HSLA–100 plate from the trial productions, an evaluation program commenced to certify HSLA–100 steel plate and weldments for use in ship structural applications. The program included the characterization of production HSLA–100 steel plate mechanical, physical, and fracture properties; evaluation of weldability and welding process limits for structures of high restraint; studies of fatigue properties and effects of marine environments on HSLA–100; and the fabrication and evaluation of large scale structural models to validate the laboratory-developed welding process parameters. The findings of the HSLA–100 steel evaluation are summarized as follows [16]:

The tensile and impact toughness requirements set for HSLA–100 Steel Plate were met in thicknesses from 1/4 to 3 3/4 inches in the commercial productions. Optimum properties in HSLA–100 plate resulted from aging temperatures from 1150° to 1275°F, with double austenitization and quench was used to develop fine grain size and optimum toughness in HSLA–100 steel plate in gages over 1 1/4 inches.

The strength and toughness of weld metals deposited by the SMAW, SAW, GMAW-P, and GMAW-S processes, using the welding consumables qualified for HY–100 welding, met requirements when welded within the process operating envelopes for HY–100. No "hard" microstructures were indicated, and the Charpy V-notch toughness of the HAZ in HSLA–100 weldments was equal to or greater than the weld metal toughness.

Strength or toughness deficiencies in some HSLA–100 weldments were due to limitations of the welding consumables, not related to welding HSLA–100 steel, per se. Similar or lower results were shown in welding HY–100 base plate with the same consumables. No HSLA–100 weldment HAZ problems were encountered.

The weldability evaluation indicated that weld metal cracking in SAW and SMAW systems limits the ability to weld HSLA–100 without preheat and a maximum interpass temperature of 60°F under highly restrained conditions. Preheat was recommended for SAW and SMAW, based on the weld metal cracking tendencies noted for these flux-assisted processes in the weldability testing.

For the GMAW process, it was concluded that acceptable weldments in HSLA–100 plate up to 2 inch gage could be fabricated using a 60° F minimum preheat and a 60° to 300° F interpass temperature range.

It was demonstrated that HSLA–100 fillet weld strengths were equivalent to HY–100 welds using the same process, filler metal, and fillet size.

Explosion bulge and crack starter explosion bulge tests of 3/4 inch thick plate and 2 inch thick GMAW, SMAW, and SAW weldments of HSLA–100 steel were conducted. The weldments exhibited no indications of hydrogen damage and passed the explosion bulge test requirements for HY–100 plate.

The results of the low-cycle fatigue crack initiation studies of HSLA–100 steel and weldments under cyclic loading in air and in marine environments showed low-cycle fatigue properties equivalent to HY–100 steel and weldments. HSLA–100 and HY–100 exhibited similar fatigue crack propagation behavior.

Both HSLA–100 and HY–100 steels showed similar high-cycle fatigue properties. High-cycle tests of HSLA–100, HY–100, and HY–80 GMAW butt weldments demonstrated that all the weldments had similar fatigue properties.

General corrosion, crevice corrosion, galvanic corrosion, high velocity seawater parallel flow and cavitation tests of HSLA–100 steel in seawater showed that the corrosion behavior of HSLA–80, HY–80, and HSLA–100 steels was comparable.

HSLA–100 plate, weld metal, and weld HAZ did not show any susceptibility to stress corrosion cracking exposed at –1000 mV at or above stress corrosion cracking threshold stress intensity values determined for HY–100 and weld metals.

The fabrication of a series of structural performance models was completed under shipyard welding conditions. These were evaluated and demonstrated satisfactory structural performance.

EXPANSION OF THE CU-STRENGTHENED STEEL SYSTEM

HSLA–80 steel was not used in ship structures in critical or heavy plate applications due to inconsistent fracture toughness in the heavier gages. The fundamental fracture process studies in HSLA–80 steel [18] showed that coarse-grained, polygonal ferrite and the local accumulation of secondary transformation products (carbide-rich islands) were deleterious to low-temperature cleavage fracture resistance.

The research indicated that a uniformly small grain size and wider distribution of small carbides would reduce the fracture transition temperature. The HSLA–100 alloy design produced a homogeneous, fine-grained bainitic microstructure, which dispersed the secondary transformation products. The same microstructural modification was the goal for alloy design of modified HSLA–80 in heavy plate gages to ensure high fracture toughness at all service temperatures.

The alloy development study was conducted to microstructurally modify HSLA–80 steel using laboratory-scale heats to study the effects of Mn, Ni, Mo, Cu, Cr, Cb and C in hot rolled, quenched and aged HSLA–80 plate. Microstructural analysis was conducted to develop composition ranges, meeting the strength and toughness requirements, where ferritic microstructures were not present.

In the production of HSLA–100 steel plate under the certification program, difficulties were experienced in keeping below the maximum yield strength in plate gages less than 1 inch, unless very high aging/tempering temperatures were used. These results suggested that a much leaner composition could meet HSLA–100 requirements in thinner plate gages. Thus, the development program for modification of HSLA–80 also included analysis to formulate a recommended chemistry range for both (1) a lower cost, intermediate composition HSLA–100 for plate gages 1 inch and less, and (2) heavy plate HSLA–80 (gages greater than 1 1/4 inch) with improved low temperature toughness. The composition ranges developed from the study are given in Table 3.

The overlapping chemistry ranges allowed for the production of a single 165–ton electric furnace melt of modified HSLA–80/100 steel, to provide slab for both lower cost HSLA–100 plate in gages less than 1 inch and heavy plate HSLA–80 with improved low temperature toughness.

An evaluation of HSLA–100 steel plate of a composition intermediate to the alloy contents of HSLA–80 and HSLA–100 was conducted which included the production of plate, characterization of the mechanical properties of production plate, screening of weldability and welding process limits, investigation of fatigue properties, and an evaluation of the explosion bulge resistance of HSLA–100 plate to the intermediate composition:

The tensile properties, Charpy V-notch impact toughness, and dynamic tear toughness of the intermediate composition HSLA–100 plates, up to 1 inch gage, inclusive, met or exceeded the requirements of MIL–S–24645A for HSLA–100 steel plate;

The weldability evaluation demonstrated the same HAZ cracking resistance exhibited by the richer composition HSLA–100 plate. For the GMAW process, acceptable weldments were fabricated in intermediate composition HSLA–100 using 60° F preheat and a 60° to 300° F interpass temperature range. Weld metal cracking in SAW and SMAW systems still limited the ability to weld HSLA–100 by these processes without preheat under highly restrained conditions;

Low- and high-cycle fatigue properties of intermediate composition HSLA–100 were similar to those of the richer composition and to HY–100;

Explosion tests of 1 inch gage GMAW butt joints met or exceeded the requirements.

In summary, the results of the evaluation indicated that the intermediate composition HSLA–100 plate in gages of 1 inch and less can meet the performance of HSLA–100 plate of the richer composition.

The steelmaking process for the richer composition HSLA–100 steel in the certification program used plate rolled from slabs which were hot rolled from ingots (about 65% yield). However, the most economical production of HSLA–100 light gage plate would use strand casting of slabs directly from the melt (about 80% yield). Strand casting of HSLA–100 had not been performed due to the risk involved with high alloy content. The intermediate composition is closer to the HSLA–80 composition, for which strand casting is typical practice. The production benefits, lower alloy content, and aging temperatures could result in plate cost of HSLA–100 steel plate to the intermediate composition being less than that of HY–100 steel.

WELDABILITY OF HSLA–80 AND HSLA–100 STEELS

The HAZ of the very low carbon, Cu-strengthened HSLA steels typically does not harden. HSLA–80 and HSLA–100 steel HAZ microstructures are less sensitive to hydrogen cracking, and thus achieve excellent weldability. The qualification programs demonstrated that HSLA–80 and HSLA–100 were significantly more resistant to hydrogen cracking than HY–80 and HY–100, such as to allow a relaxation of preheat requirements.

However, the weld metals were demonstrated to be the "weak link" in the weldability of the 100,000 psi yield strength base plate/weld metal system. Weld metals deposited by the flux-assisted welding processes (SMAW, FCAW, and SAW) were less resistant to hydrogen cracking than the GMAW process. Reduction of welding preheat requirements was achieved for some welding consumables, while other consumables such as the flux for SAW will require a specified maximum allowable moisture or hydrogen content before a reduction of welding preheat can be made. Research is in progress to develop welding consumables specifically for preheat-free welding of HSLA-100. But, the lesson learned from the HSLA-100 program was that the development of high strength steel systems should have an initial focus on weld metal and welding product development, while plate development will be the less difficult accomplishment.

HIGH PRODUCTIVITY STEEL RESEARCH AND DEVELOPMENT

Because of the emphasis on affordability in ship and submarine construction, the Navy has continued steel research and development programs on lower-cost alternatives to the HY and Cu-strengthened steel systems. The goal is to achieve reductions in the overall total of plate, fabrication, and inspection cost. These research programs include the development of thermo-mechanically controlled-processed (TMCP) steels, ULCB (ultra-low carbon, bainitic) steels, and accelerated-cooled/direct-quenched (AC/DQ) HSLA steels to meet Navy performance goals.

THERMO-MECHANICAL CONTROLLED PROCESSING

The modern metallurgical approach to high-strength, high-toughness, weldable steel plate makes use of controlled processing (microalloying, controlled rolling, and controlled cooling) [19]. Within limits of strength and plate gage, the properties can be achieved without the need for post-rolling heat treatment. The broad category which encompasses this steel plate technology is thermo-mechanical, controlled processing (TMCP).

TMCP involves conditioning the high temperature phase of steel (austenite) followed by controlled cooling to result in an optimized microstructure and fine grain size. For a given composition, the primary plate rolling factors which influence final properties are the rolling schedule and the cooling rate. In TMCP, "controlled rolling" (CR) is conducted and relies on microalloying; small amounts (0.001 to 0.5%) of elements such as niobium, vanadium, titanium, and boron; to retard or suppress recrystallization of deformed austenite [20]. Elongated austenite grains with large grain boundary areas are formed, in addition to deformation bands within the austenite, to act as nucleation sites for a fine-grained product phase. TMCP is accomplished by a careful adjustment of the heating and rolling schedule needed to reduce the thickness of the steel ingot or slab to the thickness of the final plate. The usual hot rolling practice in plate mills is performed as fast as possible without any regard to the final microstructure and grain size.

Conventional hot rolling schedules for steel plate production typically range from 2300° F roughing passes to finishing passes above 1700° F. In this range, the hot strength of the steel is low, resulting in rapid slab-to-plate reduction schedules. Grain refinement in steels is enhanced through a combination of CR and microalloying. The primary grain refinement mechanism in CR is recrystallization of austenite during hot deformation; dynamic recrystallization [21]; whereas static recovery and recrystallization take place between rolling passes and during cooling at the completion of the rolling process. The process of recrystallization is influenced by the temperature and degree of deformation which takes place during each rolling pass.

Three stages of controlled rolling have been defined [22]:

1. Deformation in the austenite recrystallization region (typically above 1800° F), where coarse austenite is refined through repeated recrystallization due to the continuous deformation that takes place. However, the total grain refinement which can be achieved is limited;
2. Deformation in the austenite non-recrystallization region (between about 1740° F and 1600° F), where the formation of deformation bands in unrecrystallized austenite provides additional nucleation sites for transformation; and
3. Deformation in the two phase austenite-ferrite region (below the transformation temperature).

A fourth stage of TMCP includes AC/DQ after CR [23]. Figure 2 shows a schematic illustration of the stages of TMCP. TMCP and controlled cooling (AC/DQ) processes are necessary for the production of the advanced high productivity steels to be subsequently discussed.

Among the candidate steels evaluated in the HSLA–80 program were controlled rolled, microalloyed steels. In plate product up to 1/2 inch, these steels met the required strength and showed good weldability; however, low temperature toughness equivalent to HY–80 steel could not be obtained.

AC/DQ PROCESSING

In the final stages of TMCP, austenite grain growth can be further suppressed by rapid cooling from the finish rolling temperature. Accelerated cooling (AC) is the cooling of plate after CR at an intermediate cooling rate, such that the processed plates can be used in the as-cooled state without tempering [24]. Plates given AC are typically cooled through the transformation temperature range (1470° to 930° F), and then air-cooled. The air cooling after AC to ambient temperature provides self-tempering.

Direct quenching (DQ), on the other hand, can not eliminate the tempering stage of processing. Plates that are direct quenched are rapidly cooled from the transition temperature range to ambient temperature (martensitic or bainitic structure), and tempering of the as-quenched microstructure is necessary to provide an optimum combination of strength and toughness. A tempering treatment following DQ allows for very fine precipitates to provide maximum strengthening.

Figure 2 schematically showed the AC/DQ processing. In general, the cooling rates for accelerated cooling are less than for direct quenching. The cooling rate to produce optimum properties is influenced by the composition and plate gage. AC is generally considered to be in the range of 9 to 36° F/s and DQ to be from 18 to 180° F/s [25]. In common, AC/DQ processing are means of controlling the austenite transformation, and cooling must begin above the transformation temperature [25,26]; thus omitting the third stage of TMCP. AC/DQ processing, however, requires CR for grain refinement. In the absence of CR, only preferential sites on equiaxed austenite grain boundaries would serve as nucleation sites in transformation and result in coarse bainitic or martensitic structures. The combination of CR and AC/DQ, total TMCP, results in a fine, uniform grain structure [24].

The types of high strength steel plate currently produced by U.S. steel mills remain based on steelmaking and processing technology where large amounts of alloying elements (chromium, nickel and molybdenum) are required, especially for thick sections. Production of high strength steel plate for military applications is labor and energy intensive, requiring multi-step heat treatment to obtain properties. Thus, the alloy steels have a relatively high strategic metal content and production energy requirements.

AC/DQ processing provides an approach for obtaining reduced alloy content in thicker sections, while maintaining good weldability and toughness. AC/DQ processing involves on-line cooling, concurrent with plate rolling, to take advantage of the conditioning of the high temperature austenite phase [27]. By forcing microalloying elements (principally niobium) to remain in solution in austenite for longer periods at higher temperatures, superior combinations of strength and toughness can be obtained at lower alloy contents. A recent investigation [17] on HSLA–80, which was direct quenched after a controlled rolling schedule, produced a steel which achieved 100,000 psi yield strength while still meeting the HSLA–80 toughness requirements in 1/2 inch-thick plate.

The Navy AC/DQ Steels Program takes advantage of "clean steel" melting practices installed by U.S. steelmakers. These include vacuum degassing equipment, ladle treatment stations, and improved basic oxygen furnace (BOF) capabilities to aid in the reduction of tramp elements (nitrogen, hydrogen, sulphur) which degrade high strength steel properties. The program includes prototype, demonstration production and qualification testing of AC/DQ steel plate. In the AC/DQ steels program, the melting and all molten metal treatments are conducted in the U.S., while plate rolling and AC/DQ processing was performed by foreign plate mills, with TMCP–AC/DQ capability and a licensing option, as partners to the U.S. steelmaker.

As the Navy had no experience with AC/DQ processed steels, an evaluation of metallurgical, mechanical, and weldability features of several commercial AC/DQ steel plates obtained from foreign sources was conducted for AC/DQ steels over a yield strength range of 65,000 to 100,000 psi. A summary of the evaluation follows:

> AC/DQ processing can achieve the strength and toughness requirements for a specific grade/thickness with reduced alloying as compared to conventionally-processed HY and HSLA steels;
>
> On the basis of microstructural examinations, all of the plates exhibited features consistent with controlled rolling practices;
>
> 2 inch gage Grade 65 plate exhibited good weldability and met all mechanical property requirements for Navy Grade 65, except for dynamic tear toughness;
>
> 1 1/2 inch gage Grade 80 plate used significantly less alloying (Cr, Ni and Mo) than either HY–80 or HSLA–80, exhibited good weldability, and met mechanical property requirements. The plate passed explosion bulge test acceptance criteria for HY–80;
>
> Grade 100 plate (1 1/2 inch gage) was very lean in alloying in comparison with either HY–100 or HSLA–100. Weldability evaluations indicated that weld metal cracking limited the ability to weld Grade 100 without preheat under highly restrained conditions.

The alloy savings potential with AC/DQ steel technology is illustrated in Table 4, where the composition of the Grade 100 plate included in the DTRC study is compared to HY–100 and HSLA–100 steels.

ULCB STEEL PLATE DEVELOPMENT

The ULCB steels had their origins in the late 1960's with studies of steels with carbon levels from 0.03 to 0.095%, heat treated to produce lower bainite or acicular ferrite. The steels had yield strength levels in the range of 40,000 to 80,000 psi and showed superior impact toughness with very low toughness transition temperature. Ford Research Laboratory produced a ULCB steel with a 130,000 psi yield strength and excellent low temperature impact toughness in *air cooled* plate of the following composition:

0.03 C, 3 Mo, 3 Ni, 0.7 Mn, 0.3 Si, 0.5 Nb

Although the properties of the steel were very impressive, the alloy was expensive, due to the high Ni + Mo content, and the ultra-low carbon content was difficult for steel mills of the 1970's to produce.

The low carbon rendered the steel resistant to cold cracking for good weldability and stimulated the concept of very low-carbon, bainitic, air-cooled steels for high strength structural applications if a more economical alloy system could be developed. Alloy design for high toughness, weldable steel for use in high strength pipelines by the Climax Molybdenum Laboratory produced a steel with a composition of 0.05 C, 1.6 Mn, 0.25 Mo, 0.05 Nb, a yield strength of 70,000 psi, and a Charpy impact transition temperature of –76° F in commercial production. Spray cooling was required to produce the acicular ferrite structure in 1/2 inch-thick plate. Further studies of the low carbon Mn-Mo-Nb steel system showed: (1) reducing the carbon reduced the toughness transition temperature with an optimum carbon content between 0.002% and 0.05%, and (2) for alloys with less than 0.05% carbon, additions of up to 2% manganese continued to increase strength and lower the fracture transition temperature.

The low carbon Mn-Mo-Nb steel alloy design was further developed in Japan [28] where the system was enhanced by microalloying with boron and titanium. The addition of a small amount of titanium (0.012 to 0.016%) increased the strength-to-toughness balance of these steels, since titanium in the form of TiN was stable to higher temperatures than Nb(CN) and increased the temperature range for austenite grain refinement from hot working. This microalloying tool and TMCP were used by Nippon Steel Corporation to develop the commercial family of steels called ultra-low carbon bainitic (ULCB) steels [29,30]. The ULCB pipeline steels were very low carbon steel with 2% manganese and microalloying additions of Nb, Ti, B, and Al. The chemical composition of the X–80 grade of ULCB steel was as follows: 0.02% C, 2.0% Mn, 0.03% Ni, 0.05% Nb, 0.02% Ti, 0.04% Al, 0.001% B. The 0.02% carbon level requires the use of expensive electrolytic manganese (in lieu of carbon-bearing, ferro-manganese) and high-quality steelmaking practices, since the composition limits and processing procedures are very stringent.

The Navy conducted studies of the Nippon Steel X–65 grade plate and domestic laboratory melts of ULCB steels to determine their potential for achieving the fracture toughness required for HY–80/100 applications. Feasibility was demonstrated for developing an air-cooled, 100,000 psi yield strength ULCB steel with high toughness in plate and HAZ. A family of ULCB steels was demonstrated on a laboratory scale that used higher amounts of Mo, Ni, and Nb with controlled TMCP to obtain high toughness and yield strengths from 80,000 to 130,000 psi [31,32].

The ULCB development program culminated in a production demonstration in which a 50–ton melt of ULCB–100 steel was produced by Allegheny-Ludlum Steel Corp. and rolled to 1 inch gage plate on a laboratory mill. The composition and properties of the ULCB–100 are compared to HY–100 and HSLA–100 steels in Table 5.

As evident in Table 5, the alloy content of the ULCB steel is rich compared to the lean compositions of the AC/DQ steels and is comparable to both HY–100 and HSLA–100. However, the metallurgical studies on ULCB steels have indicated that these compositions are potential systems for high strength welding wire electrodes. The cooling rate insensitivity of the heavy plate ULCB steels indicates that these systems can maintain strength and toughness over a broad range of welding heat input, with a high resistance to hydrogen cracking (cold cracking).

HSLA TECHNOLOGY SUMMARY

The primary productivity benefits of HSLA technology are derived from improvement in weldability characteristics. The increase in cold cracking (hydrogen cracking, delayed cracking, underbead cracking) resistance of the HSLA steels compared to their HY steel counterparts is well documented by both Navy laboratory and industry evaluations [13,33,34]. Weld metal development will be necessary in order to fully utilize the cost saving potential afforded by the use of HSLA steels.

In summary, there are demonstrated productivity benefits which are possible with HSLA steel technology. As described in the previous sections, the R&D programs in progress will result in continued improvement in the weldability of HSLA steel plate with reduction of plate cost, while maintaining properties. This is illustrated in Figure 3, the Graville weldability diagram used as Figure 1, comparing carbon equivalents for HY–100, HSLA–100, the prototype ULCB–100 and a speculated DQ–100 composition.

Affordability will be a primary consideration in future ship and submarine construction programs. Although other materials offer advantages over steel hull systems, the economic, producibility, and industrial base considerations inevitably favor continued development of high strength steel systems; namely hull fabrication by existing production yards, where the welding equipment, processes, practices, and controls are only applicable to high strength steel fabrication. In order to meet its future needs, the Navy continues research and development of high productivity, high strength steel systems for hull fabrication.

Table 1. Specified chemical compositions and mechanical properties of HY–80, HY–100, and HY–130 steels.

(Major elements for heavy gage plate, greater than 1-¼ inch)

ELEMENT (weight %)	SPECIFIED CHEMICAL COMPOSITION (maximum unless a range is shown)		
	HY-80	HY-100	HY-130
C	0.13-0.18	0.14-0.20	0.12
Mn	0.10-0.40	0.10-0.40	0.60-0.90
P	0.015	0.015	0.010
S	0.008	0.008	0.008
Si	0.15-0.38	0.15-0.38	0.15-0.35
Ni	2.50-3.50	2.75-3.50	4.75-5.25
Cr	1.40-1.80	1.40-1.80	0.40-0.70
Mo	0.35-0.60	0.35-0.60	0.30-0.65
Cu	0.25	0.25	0.25
V	0.03	0.03	0.05-0.10
Al	*	*	0.01-0.05
Cb	*	*	0.02
Ti	0.02	0.02	0.02

	MECHANICAL PROPERTIES (minimum unless a range is shown)		
Yield Strength (ksi)	80-99.5	100-120	130-150
Tensile Strength (ksi)	*	*	*
Elongation (%)	20	18	15
Reduction of Area (%)	50	45	50
Charpy Impact (ft-lb)	60 @ 0°F and 35 @ -120°F	60 @ 0°F and 40 @ -120°F	80 @ +30°F and 40 @ -120°F
Dynamic Tear (ft-lb)	450 @ -40°F	500 @ -40°F	600 @ +30°F and 500 @ -20°F

* = not specified.

Table 2. Specified chemical compositions and mechanical properties of HY–80, HSLA–80, HY–100, and HSLA–100 steels.

(Major elements for 1¼ inch gage plate)

ELEMENT (weight %)	SPECIFIED CHEMICAL COMPOSITION (maximum unless a range is shown)			
	HY-80 MIL-S-16216K	HSLA-80 MIL-S-24645A	HY-100 MIL-S-16216K	HSLA-100 MIL-S-24546A
C	0.13-0.18	0.06	0.14-0.20	0.06
Mn	0.10-0.40	0.40-0.70	0.10-0.40	0.75-1.15
P	0.015	0.020	0.015	0.020
S	0.008	0.006	0.008	0.006
Si	0.15-0.38	0.40	0.15-0.38	0.40
Ni	2.50-3.50	0.70-1.00	2.75-3.50	3.35-3.65
Cr	1.40-1.80	0.60-0.90	1.40-1.80	0.45-0.75
Mo	0.35-0.60	0.15-0.25	0.35-0.60	0.55-0.65
Cu	0.25	1.00-1.30	0.25	1.45-1.75
Cb	*	0.02-0.06	*	0.02-0.06
V	0.03	0.03	0.03	0.03
Ti	0.02	0.02	0.02	0.02
	MECHANICAL PROPERTIES (minimum unless a range is shown)			
Yield Strength (ksi)	80-99.5	80-100	100-120	100-125
Tensile Strength (ksi)	*	*	*	*
Elongation (%)	20	20	18	18
Reduction of Area (%)	50	50	45	45
Charpy Impact (ft-lb)	60 @ 0°F and 35 @ -120°F	60 @ -120°F	60 @ 0°F and 40 @ -120°F	80 @ 0°F and 60 @ -120°F
Dynamic Tear (ft-lb)	450 @ -40°F	*	500 @ -40°F	*

* = not specified.

Table 3. Chemical composition ranges for HSLA-80/100 steel plate.

ELEMENT (weight %)	SPECIFIED CHEMICAL COMPOSITION (maximum unless a range is shown)			
GRADE	HSLA-80		HSLA-100	
PLATE GAGE (inches)	1¼ and less	all gages	1 and less	all gages
	(Presently specified in MIL-S-24645A, Amendment 1) ↓		↓	↓
C	0.06	0.06	0.06	0.06
Mn	0.40-0.70	0.85-1.15	0.75-1.15	0.75-1.05
P	0.020	0.020	0.020	0.020
S	0.006	0.006	0.006	0.006
Si	0.40	0.40	0.40	0.40
Ni	0.70-1.00	1.70-2.00	1.50-2.00	3.35-3.65
Cr	0.60-0.90	0.45-0.75	0.45-0.75	0.45-0.75
Mo	0.15-0.25	0.45-0.55	0.30-0.55	0.55-0.65
Cu	1.00-1.30	1.00-1.30	1.00-1.30	1.45-1.75
Cb	0.02-0.06	0.02-0.06	0.02-0.06	0.02-0.06

Table 4. Comparison of specified chemical compositions and mechanical properties of HY–100 and HSLA–100 steels to Grade 100 direct quenched (DQ) production steel plate.

ELEMENT (weight %)	SPECIFIED CHEMICAL COMPOSITION (maximum unless a range is shown)		
	Grade 100 (DQ) (ladle composition)	HY-100 MIL-S-16216K	HSLA-100 MIL-S-24546A
C	0.04	0.14-0.20	0.06
Mn	1.44	0.10-0.40	0.75-1.15
P	0.002	0.015	0.020
S	0.001	0.008	0.006
Si	0.25	0.15-0.38	0.40
Ni	0.73	2.75-3.50	3.35-3.65
Cr	0.024	1.40-1.80	0.45-0.75
Mo	0.16	0.35-0.60	0.55-0.65
Cu	0.87	0.25	1.45-1.75
Cb	0.044	*	0.02-0.06
V	0.001	0.03	0.03
Ti	0.015	0.02	0.02
N	0.003	*	*
Heat Treatment	Control Roll & DQ Temper	Hot Roll, Austenitize, Water Quench, Temper	Hot Roll, Austenitize, Water Quench, Age Harden
	MECHANICAL PROPERTIES (minimum unless a range is shown)		
Yield Strength (ksi)	94	100-120	100-125
Tensile Strength (ksi)	103	*	*
Elongation (%)	30	18	18
Reduction of Area (%)	74	45	45
Charpy Impact (ft-lb)	112 @ 0°F 40 @ -120°F	60 @ 0°F and 40 @ -120°F	80 @ 0°F and 60 @ -120°F
Dynamic Tear (ft-lb)	515 @ -40°F	500 @ -40°F	*

* = not specified.

Table 5. Comparison of specified chemical compositions and mechanical properties of HY–100 and HSLA–100 steels to ULCB–100 production steel plate.

ELEMENT (weight %)	ULCB-100 (ladle composition)	SPECIFIED CHEMICAL COMPOSITION (maximum unless a range is shown) HY-100 MIL-S-16216K	HSLA-100 MIL-S-24546A
C	0.012	0.14-0.20	0.06
Mn	1.04	0.10-0.40	0.75-1.15
P	0.005	0.015	0.020
S	0.001	0.008	0.006
Si	0.34	0.15-0.38	0.40
Ni	3.42	2.75-3.50	3.35-3.65
Cr	0.31	1.40-1.80	0.45-0.75
Mo	1.79	0.35-0.60	0.55-0.65
Cu	0.06	0.25	1.45-1.75
Cb	0.06	*	0.02-0.06
V	nil	0.03	0.03
Ti	0.004	0.02	0.02
Heat Treatment	As Rolled, Temper	Hot Roll, Austenitize, Water Quench, Temper	Hot Roll, Austenitize, Water Quench, Age Harden
	MECHANICAL PROPERTIES		
	(average)	(minimum unless a range is shown)	
Yield Strength (ksi)	96	100-120	100-125
Tensile Strength (ksi)	104	*	*
Elongation (%)	33	18	18
Reduction of Area (%)	69	45	45
Charpy Impact (ft-lb)	90 @ 0°F 40 @ -120°F	60 @ 0°F and 40 @ -120°F	80 @ 0°F and 60 @ -120°F
Dynamic Tear (ft-lb)		500 @ -40°F	*

* = not specified.

		C	Mn	P	S	Si	Cr	Ni	Mo	Cu	Cb	V	B	C.E.*
TRADITIONAL NAVY STRUCTURAL STEELS	HY-80	.15	.25	.01	.01	.25	1.40	2.70	.40	.05	—	.01	—	0.78
	HY-100	.17	.25	.01	.01	.25	1.40	2.90	.40	.05	—	.01	—	0.81
CURRENT NAVY HSLA STEELS	HSLA-80	.04	.55	.01	.005	.30	0.70	0.90	.20	1.20	.04	—	—	0.50
	HSLA-100	.04	.90	.01	.005	.25	0.60	3.50	.60	1.60	.03	—	—	0.81

*C.E. = CARBON EQUIVALENT

CARBON EQUIVALENT

$$CE = C + \frac{Mn+Si}{6} + \frac{Ni+Cu}{15} + \frac{Cr+Mo+V}{5}$$

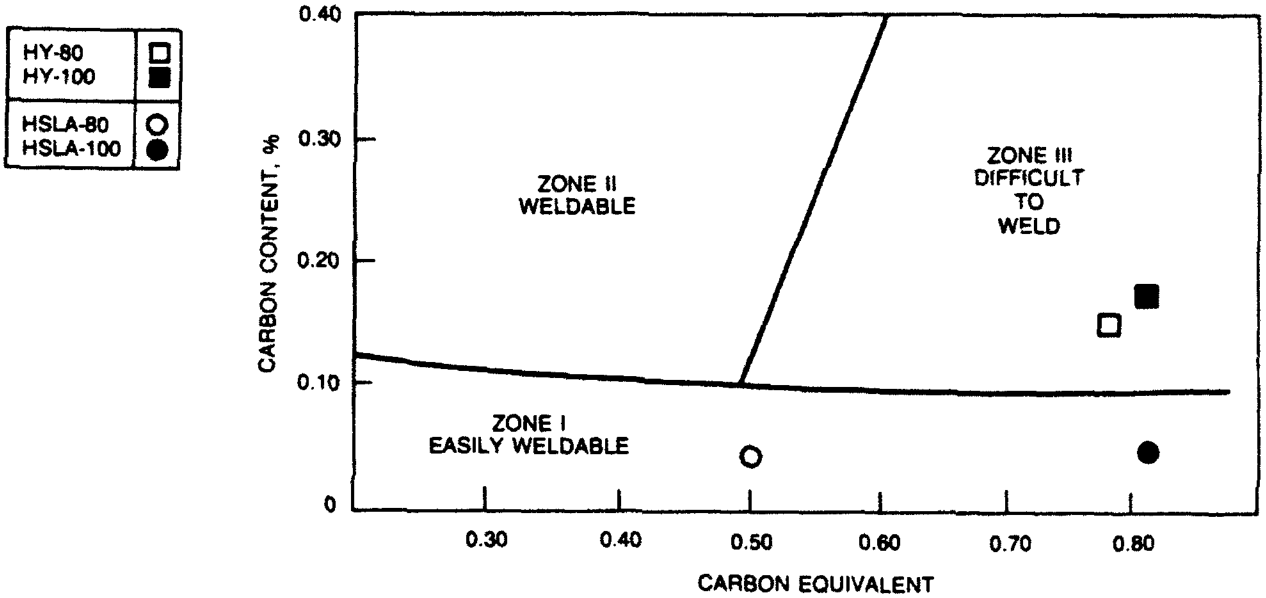
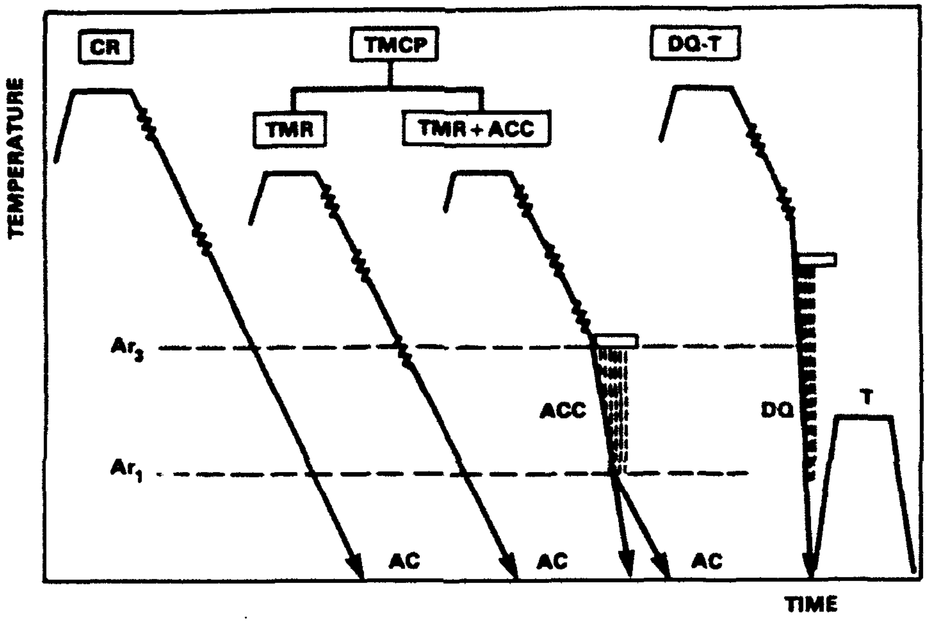


Figure 1. Chemical compositions, carbon equivalents, and weldability diagram for high strength naval steels.



CR: CONTROLLED ROLLING, TMR: THERMO-MECHANICAL ROLLING,
ACC: ACCELERATED COOLING, DQ: DIRECT QUENCHING, T: TEMPERING,
AC: AIR COOLING

Figure 2. Schematic diagram of the temperature regimes for variations of TMCP of steels.

TRADITIONAL NAVY STRUCTURAL STEEL	HY-100	0.17	0.25	0.01	0.01	0.25	1.40	2.80	0.40	0.05	–	0.01	–	■ 0.81
		C	Mn	P	S	Si	Cr	Ni	Mo	Cu	Cb	V	B	C.E.*
CURRENT NAVY HSLA STEEL	HSLA-100	0.04	0.90	0.01	0.005	0.25	0.60	3.50	0.60	1.60	0.03	–	–	● 0.81
DEVELOPMENTAL ULCB STEEL	ULCB-100	0.02	1.00	0.005	0.005	0.20	–	1.40	1.50	–	0.05	–	0.001	▲ 0.61
ADVANCED TECHNOLOGY HSLA (AC/DQ)	DQ-100	0.06	1.65	0.01	0.002	0.23	–	0.70	0.30	1.25	0.04	–	–	♦ 0.54

* CARBON EQUIVALENT

$$CE = C + \frac{Mn + Si}{6} + \frac{Ni + Cu}{15} + \frac{Cr + Mo + V}{5}$$

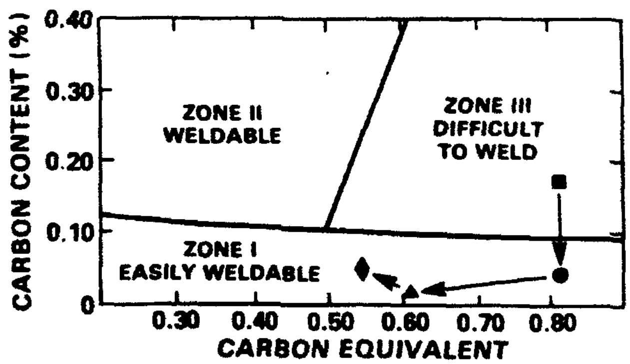


Figure 3. Chemical compositions, carbon equivalents, and weldability diagram for various 100 ksi steels.

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THE DEVELOPMENT OF HIGH-STRENGTH, COOLING-RATE INSENSITIVE

Ultra-Low-Carbon Steel Weld Metals

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ABSTRACT

High-strength steel weld metal systems have been developed based on ultra-low carbon bainitic (ULCB) metallurgy for systems in the range of 100,000 to 150,000 psi yield strength (690 to 1035 MPa). These weld metals show a potential for wire electrode formulations for HSLA steel welding with high productivity and affordability benefits in future ship hull construction. The designs of these ULCB weld metal systems and prototype welding electrodes use a bainitic metallurgy, where the as-deposited weld metal strength is independent of weld metal cooling rate with a good low-temperature toughness derived through the ultra-low carbon levels. The cooling rate independence allows the use of a wide range of welding heat input levels, while the low carbon levels reduce hydrogen-cracking sensitivity, eliminating the need for preheat and post-weld thermal soaking. The elimination of preheat, expanded welding process operational envelope, and reduced post-weld cracking risk results in higher productivity, reduced labor requirements in both time and skill, and less intensive inspection provisions for high strength steel hull fabrication.

KEYWORDS: welding, metallurgy, high-strength steels, ULCB steels, HY–80, HY–100, HSLA–80, HSLA–100.

INTRODUCTION

The welding of HY–80, HY–100, and HY–130, the high strength steels traditionally used in U.S. Navy ship construction, requires a number of fabrication controls to prevent weld cracking, which result in high fabrication costs. The Navy is conducting HSLA steel research and development programs with a goal of reducing shipbuilding costs, including the development of steel welding consumables for HSLA steel welding. This paper reviews the research conducted to develop a class of welding electrodes, based on ultra-low carbon, bainitic (ULCB) steels which demonstrate a potential for providing high strength, high toughness, and high quality weld metals for HSLA steel welding. The aim is to develop a steel welding system requiring less parameter control requirements and preheat-free welding.

HSLA–80 steel was qualified for use in ship construction after an extensive program demonstrated that the low carbon, copper precipitation-strengthened steel met the performance requirements of HY–80 steel, but was readily weldable without preheat [1,2]. Lower fabrication costs and higher productivity in construction were realized [3]. Following the HSLA–80 program, an alloy development and qualification program was conducted which resulted in HSLA–100 steel. HSLA–100 is also a low carbon, copper precipitation strengthened steel, meeting the strength and toughness of HY–100 steel, but weldable with lower preheat [4].

The primary reason for preheat in the welding of the HY-series steels was to mitigate underbead cracking (hydrogen related) in the hard, martensitic heat affected zone (HAZ). The HAZ of the very low carbon copper-strengthened HSLA–80 and HSLA–100 steels typically does not harden, and HSLA–80 and HSLA–100 steel HAZ microstructures are less sensitive to hydrogen cracking, and thus achieve excellent weldability. The welding and weldability testing demonstrated that HSLA–80 and HSLA–100 were significantly more resistant to hydrogen cracking than HY–80 and HY–100, such as to allow a relaxation of preheat requirements. However, the weld metals were demonstrated to be the "weak link" in the of the HSLA–100 base plate/weld metal system.

The welding consumables used for the qualification of HSLA–80 and HSLA–100 were the same materials already qualified for HY–80 and HY–100 welding, respectively. No separate development of welding consumables for the HSLA steels was conducted. For HSLA–100, it was found that weld metals deposited by the flux-assisted welding processes (SMAW, FCAW, and SAW) were less resistant to hydrogen cracking than the GMAW process [4]. Reduction of welding preheat requirements was achieved for some welding consumables, while other consumables such as the flux for SAW required a specified maximum allowable moisture or hydrogen content before a reduction of welding preheat could be made. It was concluded that the welding process limitations for HSLA–100 were due to the welding consumables (developed for the HY–100 system) [4,5,6] and not the weldability of the plate. The lesson learned from the HSLA–100 qualification program, regarding the development of high-strength steel systems, was that initial focus must be on weld metal and welding product development, while plate development was the less difficult accomplishment.

The HSLA steel research and development program is investigating the thermomechanical, controlled-process (TMCP) and accelerated cooled/direct quenched (AC/DQ) HSLA steels. AC/DQ processing involves on-line cooling, concurrent with plate rolling, to take advantage of the conditioning of the high temperature phases of TMCP steels. These modern HSLA steels use clean steelmaking practices, microalloying, and thermomechanical processing to retain a refined, stable, ductile, hydrogen-crack-resistant HAZ, suitable for high heat input welding. In the coming decades, it can be anticipated that modular ship hull fabrication will continue using conventional metal-arc welding processes. In order to gain higher productivity with these processes, weld metals must be developed to allow the full benefit of these advantages.

In practice, shipyards can efficiently use steel systems which can weld plate gages from 1/4 to 4 inches with a broad heat input range of 30 to 100 kJ/inch, interpass temperatures of 300° F, without deleterious effects to the weld metal. It is desirable to develop a less cooling rate sensitive weld metal deposit for welding the higher strength steels (above 690 MPa yield strength) to allow higher productivity with the conventional arc welding processes.

HIGH-STRENGTH STEEL WELDING PRODUCTS

Welding consumables have a major impact on the welding costs incurred during shipbuilding. The material cost of the consumables are not nearly as great as the costs associated with using them in welding. These costs include the expense of preheat, electrode/flux conditioning and storage, high labor costs due to productivity limitations (heat input and welding position restrictions), necessity for repair welding, and post-weld soaking to eliminate cracking, when required.

The objective in developing welding consumables for a given process is to economically produce weldments that have properties at least as good as the base material. Typically, the as-deposited weld metal strength is required to be greater than the minimum required for the base metal, with the highest toughness that can be achieved. Economy demands that, in the construction of such a massive, complex structure as a ship's hull, these properties be achieved "as welded." The difficulties in welding consumables development for high-strength steels lie in the following [7]:

(1) The welding process is an incremental melting and casting process which cannot benefit from the working and heat treatment used to obtain strength, ductility, and toughness in wrought steel products;

(2) Because it is a casting, the weld metal will inevitably contain porosity, flaws, defects, etc., to some level which cannot be removed or repaired; and

(3) The arc welding processes involve many factors that directly influence the properties of the weld metal, but cannot be measured or controlled to an accurate or precise degree.

Welding involves a complex interplay of solid, liquid, gaseous, and plasma-state phenomena, such that research in welding encompasses many scientific disciplines. Much of the research and development in welding equipment, electrodes, filler metals, fluxes, and shielding gases, has been conducted by the consumables manufacturers on an empirical basis. New product developments have primarily been evolutionary, not innovative or the result of new fundamental insights. The current electrodes designed for use with the HY–series, and employed for HSLA–80/100, represent the most highly developed welding consumables available from industrial research, with toughness superior to those produced for the mass commercial market. The specified requirements for GMAW/SAW/

GTAW wire electrodes are given in Table 1. Weld metal development for higher strength (100 ksi and above) HSLA steels is necessary to fully utilize the cost saving potential of HSLA steels. Exploratory development in high strength steel welding consumables is in progress at Navy, university, and industrial laboratories [8].

WELD METAL PROPERTY REQUIREMENTS

The philosophy of requiring weld metal yield strength to overmatch plate yield strength for submarine structures was primarily based on explosion bulge testing of mild steel weldments in the early 1950's. These experiments [9] showed that under dynamic plastic deformation, low strains developed in the weld metal versus the parent plate in proportion to the degree of overmatch in yield strength. Similar behavior was demonstrated in early HY–80 weldment explosion bulge tests. Since it was known that the fracture toughness and transition behavior of high strength weld metals was generally inferior to that of the parent plate, the overmatching requirement was a concept for "protecting" the weld metal [10].

Thus, for higher strength steels, the overmatch requirement is accompanied by a decrease in weld metal toughness and an increased propensity for weld metal cracking. In general, an increase in weld metal cooling rate (low heat input) is needed to increase yield strength to meet an "overmatching" strength requirement, and a decrease in toughness typically accompanies the increase in strength.

An increase in heat input and the concurrent increase in deposition rate reduces the overall cost of welding steels. In addition, rework and repair costs might be improved, since joint deficiencies, such as lack of sidewall fusion and interbead fusion defects occur more frequently when low heat inputs, to obtain a cooling rate to meet conditions for overmatching weld metal strength, are employed. Higher heat inputs promote better fusion characteristics, particularly in thinner section welding applications (less than 1 inch gage), where lower heat inputs are specified in order to meet minimum mechanical property requirements.

ULCB STEEL PLATE DEVELOPMENT

The ULCB steels were a result of research in the late 1960's, where steels with low carbon levels, 0.03 to 0.095%, were heat treated to produce lower bainite or acicular ferrite. The steels had yield strengths in the range of 40,000 to 80,000 psi and showed superior impact toughness with very low toughness transition temperature. Ford Research Laboratory produced a ULCB steel with the following composition: 0.03 C, 3 Mo, 3 Ni, 0.7 Mn, 0.3 Si, 0.5 Nb, which had a 130,000 psi yield strength and excellent low-temperature impact toughness in hot-rolled, *air cooled* plate. However, due to the high Ni + Mo content, the alloy was very expensive and the ultra-low carbon content was difficult for steel mills of that era to produce.

The ultra-low carbon rendered the ULCB steel resistant to cold cracking for good weldability. Alloy design for high toughness, weldable steels for high-strength pipelines by the Climax Molybdenum Laboratory resulted in a low-carbon, Mn-Mo-Nb steel system which showed that: (1) reducing the carbon reduced the toughness transition temperature with an optimum carbon content between 0.002% and 0.05%, and (2) for alloys with less than 0.05% carbon, additions of up to 2% manganese continued to increase strength and lower the fracture transition temperature.

The Navy program on ULCB steels initiated with studies of ULCB plate steels. Feasibility was demonstrated for developing an air-cooled, 100,000 psi yield strength ULCB steel with high toughness in plate and HAZ. A experimental family of ULCB steels was demonstrated on a laboratory scale that used higher amounts of Mo, Ni, and Nb with controlled TMCP to obtain high toughness and yield strengths from 80,000 to 130,000 psi [11, 12]. The most important metallurgical aspects of the ULCB plate studies were that the bainitic steels were cooling-rate independent, i.e. properties in the plates were independent of plate thickness. Also, the strength levels were directly related to composition, which could be formulated by simple formulae.

In the final stage of ULCB development program, a production demonstration, 50–ton melt of ULCB–100 steel was produced and rolled to 1–inch gage plate on a laboratory mill. The composition and properties of the ULCB–100 are compared to HY–100 and HSLA–100 steels in Table 2. As shown, the alloy content of the ULCB–100 steel was comparable to both HY–100 and HSLA–100 in total alloy content.

However, the metallurgical research on ULCB steels indicated that these compositions are potential systems for high strength welding wire electrodes. The cooling rate insensitivity of the heavy plate ULCB steels showed that the ULCB systems can maintain strength and toughness over a broad range of welding heat input, with a high resistance to hydrogen cracking (cold cracking).

ULCB WELDING ELECTRODE DEVELOPMENT

The cooling-rate insensitive bainitic microstructure of the ULCB plate steel studies, the 0.02C–Mn-Ni-Mo-Nb system, provided a basis for weld metal system with the following features:

> Weld metal deposits with cooling-rate insensitive, bainitic microstuctures for consistent strength under all welding process parameters;
>
> Ultra-low carbon for hydrogen-cracking (cold cracking) resistance and high toughness; and
>
> Fine, stable inclusions for grain size control during weld metal solidification and during multipass welding.

The alloy development research initially used GTAW of ULCB plate with deep V-grooves with matching composition electrode to determine the effects of alloying on strength, cooling rate sensitivity, and impact toughness. The results were then used to guide the optimization of the system to indicate compositions for melts of ULCB steels to be drawn to wire electrode for GMAW.

Approximately 45 composition variations within the 0.02C–Mn–Ni–Mo–Nb system were included in the initial study. The GTAW deposits in the groove were machines to provide samples for tensile and Charpy V-notch impact tests. The results of these tests indicated the following:

> Yield and tensile strength was dependent on composition by a complex function of all the alloying elements, *not* solely a function of carbon content (0.015 to 0.030%);

High Ni content can increase strength, but reaches a maximum effectiveness at 3.5 Ni with high molybdenum content (3.5 Mo);

Ni content greater than 3.5% continues to increase strength when Mo is greater than 2.5%;

Manganese demonstrated a greater than predicted effect in strength increase, but appeared to have a maximum effect at 1.5 Mn in deposits with 3.5% Mo;

The effect of Mo content on strength was less than predicted and the effect was further diminished at high Mn content;

Nb content up to 0.05% had a strong effect on strength.

The results of the ULCB composition studies are summarized in Table 3, which shows the effectiveness of each alloying element on yield strength increase (in ksi per % alloy) for two welding conditions; 60 and 120 kJ/inch average heat input (approximately 2.5 and 5 kJ/mm). These are compared to predicted effectiveness of alloying elements (strength vectors) for alloy steel plate design [13].

Figure 1 illustrates the results of all-weld-metal tensile tests of one of the experimental ULCB–100 GTAW series of the following composition:

0.018 C / 1.40 Mn / 3.48 Ni / 3.50 Mo / 0.059 Nb

In Figure 1, the results for the ULCB weld metal are compared to those for MIL–120S GMAW weld metal properties over the same range of high heat inputs. Of note is the demonstration of the insensitivity of the ULCB weld metal strength to cooling rate (welding heat input). The effect of multipass welding on the ULCB weld metal strength was simulated by the "Gleeble" apparatus, which imposes a programmed thermal cycle on a metal sample by controlled resistance heating. In these tests, an all-weld metal sample of the ULCB deposit was subjected to one or more time/temperature cycles representing multiple passes deposited under the same welding parameters as the initial weld. Tensile tests were conducted after the multipass simulation. Figure 2 shows the results for two of the compositions, including the weld metal used in Figure 1. In all cases the weld metal strength remained stable or slightly increased.

It is well known that carbon content, grain size, and inclusions are significant metallurgical factors controlling the toughness of steels; both transition temperature and upper shelf toughness. Figure 3 shows Charpy V-notch impact toughness results for several of the ULCB GTAW weld metal deposits. In only a few cases did the toughness exceed the goal of 60 ft-lb at 0° F and 45 ft-lb at –60° F. In most cases, the reduced low-temperature toughness was due to coarse MnS or oxides in the deposit. Although the experimental melts were vacuum induction melted, inclusions in the original plates persisted in the GTAW remelting and deposition.

Studies in progress to produce ULCB wire electrode for GMAW used vacuum arc remelting (VAR) for improved cleanliness for reducing the effects of coarse inclusions on weld metal toughness. Included in the continuing program are studies of the effects of microalloying to promote small, uniform distributions of inclusions for grain size control (toughness increase), and the effects of shielding gas composition on the transfer of reactive elements in the ULCB wire into the weld deposit. The compositions selected for the VAR melts, based on statistical analysis of the GTAW weld deposit results were variations of the following system: C – 2 Mn – 3 Mo – 5 Ni. The composition of melts for the optimization of toughness were as follows:

Effect of Carbon Content

0.02 C – 2 Mn – 3 Mo – 5 Ni

0.04 C – 2 Mn – 3 Mo – 5 Ni

0.06 C – 2 Mn – 3 Mo – 5 Ni

Effect of Microalloy Inclusions

0.02 C – 2 Mn – 2.5 Mo – 5 Ni – 0.006 N – 0.010 Ti

0.02 C – 2 Mn – 2.5 Mo – 5 Ni – 0.006 N – 0.018 Ti

0.02 C – 2 Mn – 2.5 Mo – 5 Ni – 0.006 N – 0.022 Ti

These ULCB steel ingots were sawcut and surface conditioned were rolled to rod for wire drawing. The rods were progressively drawn, stress relieved, cleaned, and coiled for GMAW 1/16–inch diameter wire electrode.

SUMMARY

High-strength steel weld metal systems have been developed based on ultra-low carbon bainitic (ULCB) steel metallurgy. The designs of these ULCB weld metal systems and prototype welding electrodes use a bainitic metallurgy, where the as-deposited weld metal strength is independent of weld metal cooling rate, stable for multipass welding. Good low-temperature toughness was derived through the ultra-low carbon levels and fine microalloy inclusions to control grain size. The cooling rate independence allows the use of a wide range of welding heat input levels, while the low carbon levels reduce hydrogen-cracking sensitivity, eliminating the need for preheat and post weld thermal soaking.

Table 1. Specified chemical compositions and mechanical properties for GMAW/SAW/GTAW wire electrodes, MIL–XXXS type, for welding the HY–series steels.

ELECTRODE SPECIFICATION MIL-E-	23765/2D	23765/2D	24355B
ELECTRODE TYPE MIL-	100S-1	120S-1	140S-1
FOR WELDING	HY-80	HY-100	HY-130
ELEMENT (weight %)	SPECIFIED CHEMICAL COMPOSITION (maximum unless a range is shown)		
C	0.08	0.09	0.12
Mn	1.25-1.80	0.90-2.35	1.50-2.00
P	0.012	0.012	0.010
S	0.008	0.008	0.010
Si	0.20-0.55	0.60	0.30-0.50
Ni	1.40-2.10	1.00-3.00	1.95-3.10
Cr	0.30	0.80	0.65-1.05
Mo	0.25-0.55	0.30-1.00	0.40-1.00
Cu	*	*	0.15
V	0.05	0.03	0.04
Al	0.10	0.10	0.04
Zr	0.10	0.10	0.04
Ti	0.10	0.10	0.04
	WELD METAL MECHANICAL PROPERTIES (minimum unless a range is shown)		
Yield Strength (ksi)	82-110	102-122	135-150
Tensile Strength (ksi)	*	*	*
Elongation (%)	16	14	14
Charpy Impact (ft-lb)	60 @ 0°F and 35 @ -60°F	60 @ 0°F and 45 @ -60°F	*
Dynamic Tear (ft-lb)	450 @ +30°F and 300 @ -20°F	575 @ +30°F and 400 @ -20°F	475 @ +30°F

* = not specified.

Table 2. Comparison of specified chemical compositions and mechanical properties of HY–100 and HSLA–100 steels to ULCB–100 production steel plate.

ELEMENT (weight %)	ULCB-100 (ladle composition)	SPECIFIED CHEMICAL COMPOSITION (maximum unless a range is shown) HY-100 MIL-S-16216K	HSLA-100 MIL-S-24546A
C	0.012	0.14-0.20	0.06
Mn	1.04	0.10-0.40	0.75-1.15
P	0.005	0.015	0.020
S	0.001	0.008	0.006
Si	0.34	0.15-0.38	0.40
Ni	3.42	2.75-3.50	3.35-3.65
Cr	0.31	1.40-1.80	0.45-0.75
Mo	1.79	0.35-0.60	0.55-0.65
Cu	0.06	0.25	1.45-1.75
Cb	0.06	*	0.02-0.06
V	nil	0.03	0.03
Ti	0.004	0.02	0.02
Heat Treatment	As Rolled, Temper	Hot Roll, Austenitize, Water Quench, Temper	Hot Roll, Austenitize, Water Quench, Age Harden
		MECHANICAL PROPERTIES	
	(average)	(minimum unless a range is shown)	
Yield Strength (ksi)	96	100-120	100-125
Tensile Strength (ksi)	104	*	*
Elongation (%)	33	18	18
Reduction of Area (%)	69	45	45
Charpy Impact (ft-lb)	90 @ 0°F 40 @ -120°F	60 @ 0°F and 40 @ -120°F	80 @ 0°F and 60 @ -120°F
Dynamic Tear (ft-lb)		500 @ -40°F	*

* = not specified.

Table 3. Change in yield strength (ksi) per % alloying element for experimental ULCB steels.

Alloy	< 3.5% Ni		> 3.5% Ni		Molybdenum		Carbon		< 1.5% Mn		> 1.5% Mn		< 0.45% Nb		> 0.45% Nb	
Heat Input (kJ/in)	60	120	60	120	60	120	60	120	60	120	60	120	60	120	60	120
2.5% Mo	6.25	4.17	-2.3	-1.1	—	—	76.9	230	22.5	18.4			261	130		
3.5% Mo	4.52	9.03	4.52	9.03	—	—	875	625	30.6	21.3	1.3	-1.3	267	133	467	333
2.5% Ni	—	—	—	—	4.72	4.72										
3.5% Ni	—	—	—	—	-1.0	2.0							140	70		
2.0% Mn					-3.1	-2.1			—	—	—	—				
Heuschkel	3.9	3.9	3.9	3.9	17.6	17.6	115	115	15.3	15.3	15.3	15.3	—	—	—	—
Pickering	1.1	1.1	1.1	1.1	17.4	17.4	247	247	12.0	12.0	12.0	12.0	168	168	168	168

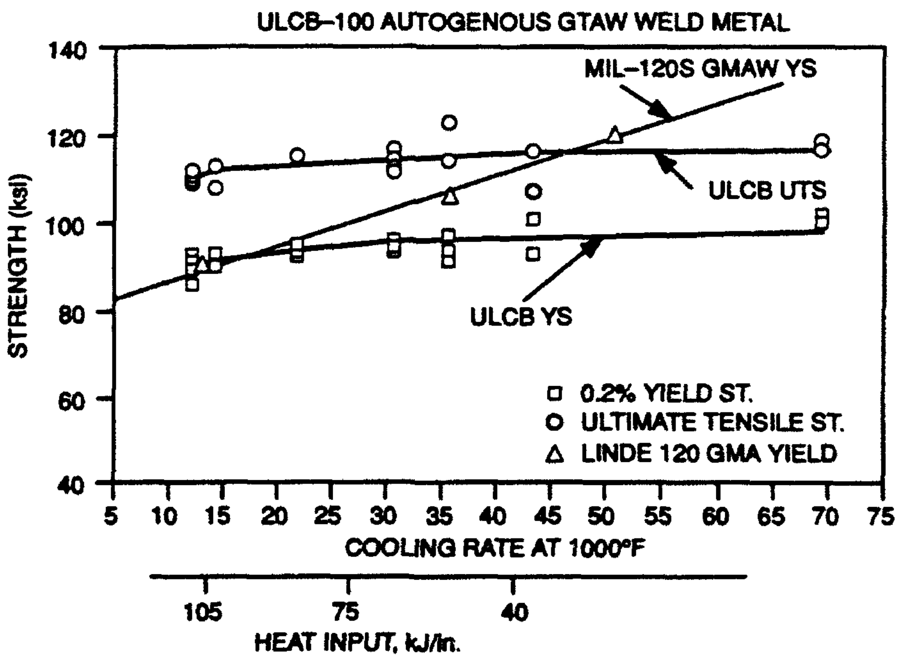

Figure 1. Strength of as-deposited ULCB weld metal versus cooling rate (welding heat input).

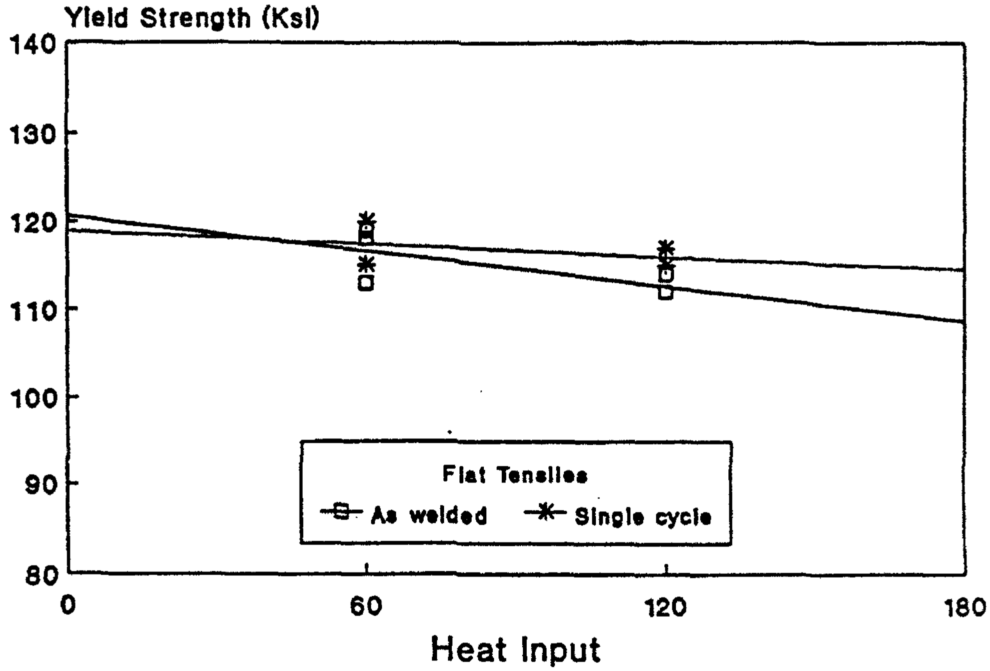

Figure 2a. Effect of simulated multipass welding thermal cycles on ULCB weld metal yield strength. 0.02 c – 1.5 Mn–4.5 Ni

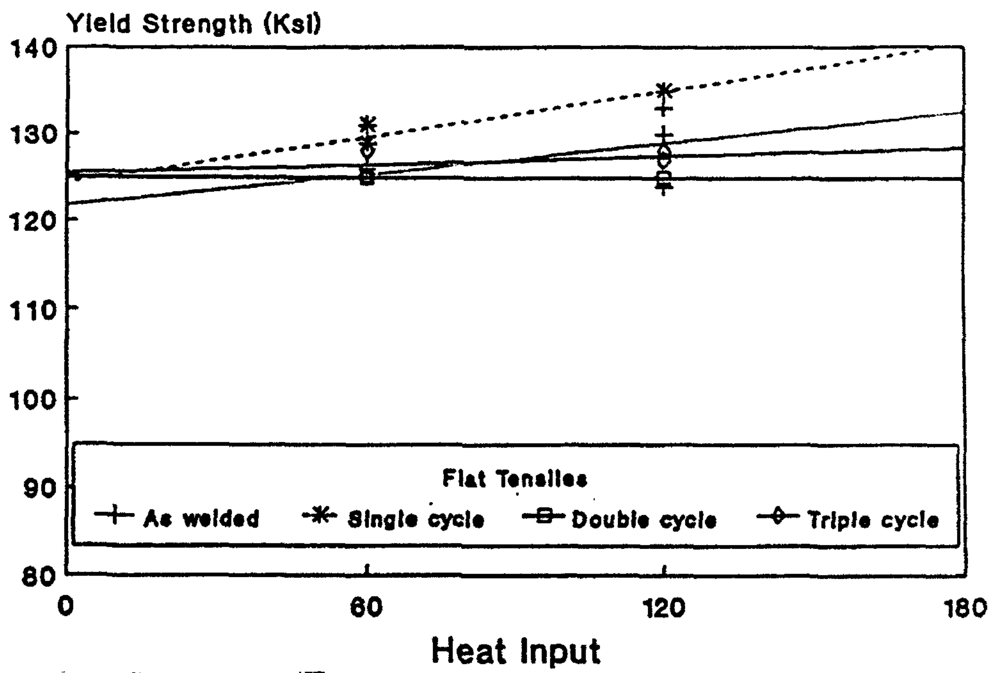

Figure 2b. Effect of simulated multipass welding thermal cycles on ULCB weld metal yield strength. 0.02 c – 1.5 Mn – 3.5 Mo – 4.0 Ni

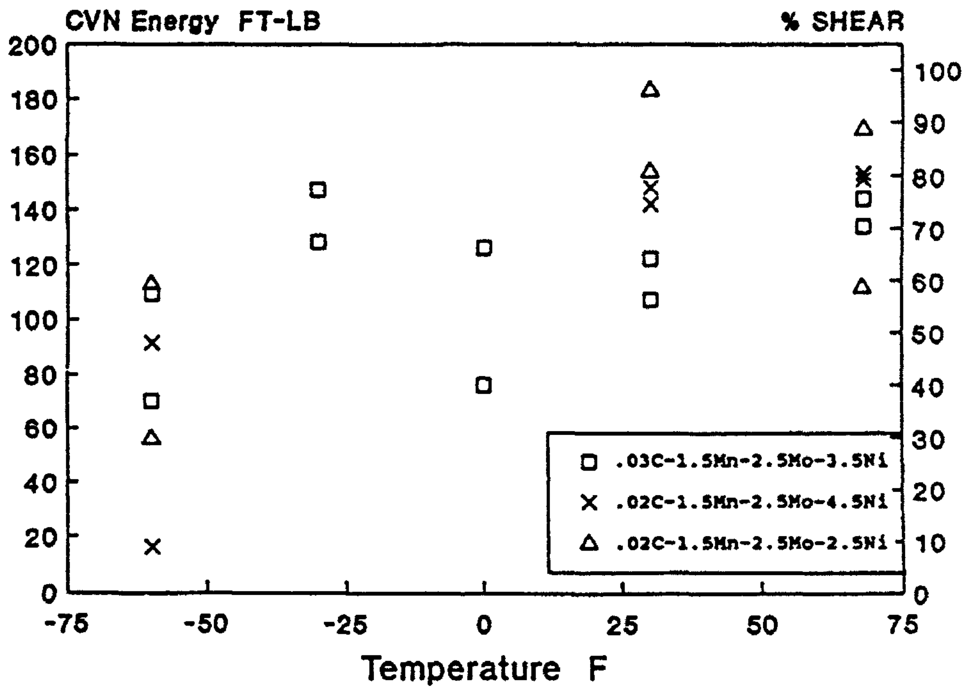

Figure 3. Charpy V-notch impact energy for ULCB weld metals.
(60 to 100 kJ/inch)

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13. ABSTRACT *(Maximum 200 words)*

Carderock Division, Naval Surface Warfare Center participated in the Indo-US Pacific Rim Workshop on "Advances in Low Carbon, High Strength Ferrous Alloys" held in New Delhi, India, 25 to 28 March 1992. E.J. Czyryca and M.G. Vassilaros gave invited presentations on the status of Navy programs on steel plate and weld metal research and development, respectively. The workshop was jointly sponsored by Office of Naval Research, Naval Research Laboratory, US Army Research Office – Far East, and the National Metallurgical Laboratory of India.

This report contains papers based on the two presentations. Paper 1 – "Advances in High Strength Steel Technology for Naval Hull Construction," by E.J. Czyryca, was presented by Mr. Czyryca. It covers an overview of US Navy steel plate research programs. Paper 2 – "The Development of High-Strength, Cooling-Rate Insensitive Ultra-Low-Carbon Weld Metals," by M.G. Vassilaros and E.J. Czyryca, was presented by Dr. Vassilaros. It presents the status of research in high-strength steel weld metal systems based on ultra-low carbon bainitic (ULCB) metallurgy. ULCB weld metals show a potential for wire electrode formulations for HSLA steel welding, where the as-deposited weld metal strength is independent of weld metal cooling rate with a good low-temperature toughness.

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